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PB-264 156

Textures in Heavily Deformed Copper

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Oct 75

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PB 264 156

Report
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MELBOURNE VICTORIA

Materials Report 99

TEXTURES IN HEAVILY DEFORMED COPPER

by

R. B. NETHERCOTT and R. A. COYLE

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SUMMARY

An X-ray study of the texture produced by wire drawing of pure copper indicated that a double <100> + <111> texture is always found. At a deformation temperature of 293K, the predominance of the <111> texture is ascribed to the dominance of recovery over recrystallisation processes, thereby maintaining the <111> deformation texture. At deformation temperatures of 373K and 473K, the <100> texture steadily increases and the <111> texture progressively decreases. This change is ascribed to the growing dominance of recrystallisation into the cube orientation over recovery. Anisotropy of elastic moduli is used to explain the preference for the cube orientation to form during recrystallisation.

The perfection of formation of the two texture components was investigated as a function of both deformation and depth below the wire surface. The <111> component is always more closely aligned with the wire axis than the <100> component. There was no evidence that a conical texture is formed during drawing.

The textures formed are neither ideally fibre nor completely cylindrical in type, but appear to be intermediate in character between these types.

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1. INTRODUCTION

Heavy cold-drawing of face-centred-cubic (f.c.c.) metals produces crystallographic textures in which the [100], [111] or [115] direction of each crystallite or cell is aligned with the direction of drawing[1,2]. These deformation textures are usually thought to have no specific crystallographic direction preferentially oriented parallel to the wire radius and are referred to as fibre textures (Fig. 1 (a)). Two other kinds of texturing in wires have recently been proposed. In the first[3], a particular crystallographic direction is oriented radially, as well as the preferred direction along the wire axis; this texture is referred to as a "cyclic" or cylindrical texture (Fig. 1 (b)). The second proposal is based on observations on extruded aluminium and aluminium-alloy rods[4]; here the individual crystallites which form the wire all have very similar orientation, as in a mosiac single crystal (Fig. 1 (c)).

Relatively little work has been reported on the textures of f.c.c. wires; most reports concentrate on the textures formed in sheets during deformation by rolling. Consideration of the rolling textures in different f.c.c. metals has led to the recognition of two common types of textures. The first, which is appropriate for f.c.c. metals and alloys of lower stacking-fault-energy, e.g. Au, Ag, brass, is designated "brass-type". The second, which is found in f.c.c. metals of higher stacking-fault-energy, e.g. Cu and Al, is designated "copper-type". In wires, these classifications correspond to the <100> and <111> fibre textures respectively.

Knowledge of wire textures is now tending to form a coherent picture but results of several workers still seem inconsistent with the usually accepted texture patterns, e.g. Freda and Cullity[1] reported increasing <111> in silver as deformation increased, whereas Ahlborn and Wassermann[5,6] found up to 100% <100> in silver for similar deformation, and English and Chin found 90% <100>[7]. Another example of variation from worker to worker is the percentage <100> texture in copper, which reportedly varies between 10 and 95%[8]. Some of these differences may be explained by the strong dependence of final texture on the initial textures present in the material[1].

In drawn copper, the fibre texture is usually duplex, i.e. two crystallographic directions, [100] and [111], are preferred along the wire axis. The relative amounts of material in each texture and the perfection of alignment of the crystallite directions depend on the conditions of drawing and the characteristics of the material.

The work reported here, dealing with textures, forms part of a wider study of the processes occurring in copper during heavy deformation and the effects of this deformation on tensile characteristics. The remainder of the study will be reported separately. This report deals with work undertaken (i) to examine the effects of strain and temperature of deformation on the drawing textures in copper, and (ii) to determine the type and perfection of the textures so formed. The report includes consideration of the mechanisms responsible for the textures present in copper.

2. MECHANISMS OF TEXTURE FORMATION IN F.C.C. METALS

The particular deformation mechanisms responsible for the crystallographic rotations and/or reorientations which form textures during or after heavy working are not known with certainty, although several have been postulated[9]. For instance, during crystallographic slip, the most densely packed planes rotate into positions in which the crystal offers its maximum resistance to shear. Sree Harsha and Cullity[10], using copper single crystals, found that, after 96% reduction in area, the texture depended on the initial orientation of the crystal. They found that the orientations of single crystals in a band between [111] and [110] in the stereographic triangle rotated towards [111] during deformation; a much wider spread of initial orientations rotated towards [100]. Similar results are also reported by Ahlborn[11].

Of the mechanisms responsible for reorientation, twinning and slip-related processes are discussed below. This discussion is followed by consideration of the general correlation between stacking-fault energy and texture in f.c.c. metals. The section concludes by considering the roles of thermal recovery and recrystallization in formation of observed textures.

2.1 Mechanical Twinning

Twinning during deformation can convert <111> textures to <115> textures. Deformation of the twinned structure by slip can then result in a <100> texture[12]. These textures are observed

in brass and silver in a temperature range consistent with the temperature-dependence of twinning in f.c.c. single crystals[12,13]; in rolling of f.c.c. sheet metals, (112) [11$\bar{1}$] twinning leads to a (552) [11$\bar{5}$] texture and then slip gives a (001) [21$\bar{1}$] texture[14].

2.2 Cross Slip

The effect of cross-slip on texture is determined by the ease with which cross-slip occurs; the more easily dislocations can move, the more completely will the metal achieve a more stable (lower energy) state. Utilising the cross-slip activation analysis of Seeger[15], Brown[16] has shown that higher temperatures and slower rates of deformation favour increasing <111>, because of the increasing ease of cross-slip as the temperature rises and as the deformation rate decreases. He showed that, on purely statistical grounds, if cross-slip is a dominant deformation mechanism in f.c.c. metals, the resulting texture should be approximately 80% <111>; the rate at which this percentage is achieved increases as the stacking-fault-energy (s.f.e.) increases.

Dillamore and Roberts[17] studied the effect of cross-slip on the deformation textures of rolled f.c.c. metals. They reported that rolling initially produces a (110)[$\bar{1}$12] texture (brass type). This texture, after further deformation, transforms by cross-slip into a (112) [11$\bar{1}$] texture (copper type). Since cross-slip usually occurs more readily in metals of higher s.f.e., the (112)[11$\bar{1}$]texture (copper type) is expected to predominate only in metals such as copper and aluminium.

One problem in the theory of texture formation by cross-slip is that cross-slip is also observed in brass alloys of low s.f.e.[18], leading to the possibility of a copper-type texture. However, given enough stress, cross-slip can occur even at low s.f.e., so that the occurrence of cross-slip is not in itself sufficient to explain the observed differences in texture; the frequency of cross-slip and the dislocation configuration must also be considered when explaining texture formation. This aspect of texture formation is relevant when considering the rolling texture in silver[19]. The texture changes from brass to copper type as the deformation temperature increases. There is a general correlation between the texture transition and the change in stacking-fault-frequency as a function of deformation temperature. The brass-type texture is associated with high stacking-fault-frequencies whereas the copper type texture is associated with low stacking-fault-frequencies. The change in texture produced by varying the deformation temperature can be interpreted by either a change in stacking-fault-energy, or in dislocation density, or both[19].

2.3 Dependence of Texture on Stacking Fault Energy

English and Chin[7] provided the most reliable study of the dependence of wire textures on s.f.e., (Fig. 2). They found that the percentage of <100> texture increased as s.f.e. decreased, from predominantly <111> (90% for Al) to predominantly <100> (90% for Ag). This change was attributed to the decreasing tendency for <111> formation by cross-slip as the s.f.e. decreases, although (as pointed out earlier) cross-slip traces have been observed in low s.f.e. metals. A more likely explanation of the increased <100> is the increased incidence of twinning at low s.f.e.; here, twinning gives a <115> texture, and then deformation of the twins by slip to <100> texture causes an increase in the total <100> present. The decrease in <100> at very low s.f.e. may be caused by the thin deformation twins, which are lamellar in shape and thus highly constrained with respect to deformation by slip. The lamellar twins may inhibit cross-slip into the <100> orientation[13]. Grewen and Wassermann[13] consider that the decrease in <100> at very low s.f.e. is not as marked as that found by English and Chin and that the combined amounts of <115> + <100> remains relatively high.

One problem in trying to correlate various results in the literature is the variation in textures observed in the same metal[8]. The information regarding experimental conditions is usually insufficient to enable an accurate comparison of results from different authors. A systematic study of texture formation is therefore required before any discussion of mechanisms can be undertaken seriously.

2.4 Softening Processes

All metals undergoing heavy deformation are in a dynamic equilibrium in which hardening and softening processes compete with each other. The softening processes, viz. recovery and recrystallisation, need to be considered when discussing textures. It has been shown[20] that pure copper and other f.c.c. metals recrystallise after heavy deformation with a pronounced cube texture (100) [001] which, in the case of wires, corresponds to a <100> texture along the wire axis. However, recovery allows the metal to soften without large orientation changes and thereby maintains the existing deformation texture.

Recrystallisation of deformed copper has recently been investigated using High Voltage Electron Microscopy[20,21]. This work shows that, prior to the nucleation of recrystallised grains, extensive sub-grain growth occurs by boundary migration. Recrystallised grains initiate in the deformed matrix and then grow into the sub-grain structure. Nuclei of several orientations form but those of cube orientation (<100>) grow more rapidly and absorb the other orientations during growth.

It is reported[22] that Beck and Gauthier have shown that rolled aluminium tends to recover fully at an homologous temperature (T/T_m) of 0.66, maintaining its rolling texture rather than recrystallising into the predominantly cube texture found after recrystallisation at higher temperatures[23]. Blade[24], while not agreeing entirely with Beck and Gauthier regarding the temperature of recrystallisation into cube orientation, does show that aluminium will recover and maintain its deformation texture at $T/T_m < 0.56$.

3. EXPERIMENTAL

High conductivity (oxygen-free) copper (>99.99%) wires were drawn from annealed 5.7mm diameter rod to various diameters down to 0.18mm, i.e. strains (ϵ) up to 7. The drawing speed was 87mm/sec and the strains per pass were 0.25 ± 0.01 for total strains less than 4.8, 0.16 ± 0.02 for strains from 4.8 to 6.4 and 0.25 ± 0.02 for strains from 6.4 to 7. Colloidal graphite was used as a dry lubricant. The temperatures of drawing were obtained as follows:

(i) room temperature (293K),
(ii) a boiling-water bath (373K), and
(iii) a long nichrome-wound furnace (473K).

The calculated increase in internal temperature of the wire due to drawing, using the formulae of Hoggart[25,26] was less than approximately 30K, and usually less than 20K.

Following other workers[6,27], it was found necessary, prior to X-ray determination of preferred orientation, to remove heavily deformed surface layers. The diameter of the wire was therefore reduced by electropolishing by 10% or 0.025mm, whichever was greater.

X-ray diffraction measurements were taken using a four-circle goniometer, which allowed scanning of the (200) and (220) reflections relating to the <100> and <111> fibre textures (Fig. 3). These reflections were equatorial or near-equatorial, thus minimising geometric corrections; corrections for reflection multiplicity, scattering and absorption were also made.

The measurements taken consisted of:

(a) the integrated intensities of the (200) and (220) reflections, which enabled the relative amounts of material in the <100> and <111> textures to be calculated, and the presence of <115> texture to be detected,

(b) the variation in arc length ($\Delta\phi$), (Fig. 4) enabling estimation of the spatial distribution within each texture.

(c) the angle of displacement (α) along the Debye-Scherrer ring of the diffraction peak from the position it would occupy if the fibre axis was accurately aligned with the wire axis.

Each of the above measurements was made on the (200) and (220) reflections for wires with strains from 0 to 7, and temperatures of deformation of 293K, 373K, and 473K.

4. RESULTS

4.1 Perfection of Texture Orientation

In drawn wires, the perfection of texturing varies with strain and may vary across the wire diameter. This tendency can result in two modifications of the texture viz.

(i) Conical texture,
(ii) Increasing alignment of texture.

(i) Conical Texture

A small displacement of a diffraction peak from its usual position on the Debye-Scherrer ring indicates that the textured directions have assumed a conical configuration about the drawing direction (Fig. 5). The angular displacement (α) of the peak approximates to the angle of inclination of the textured directions to the wire axis.

Measurements of α on copper drawn at temperatures of 293K and 473K to a strain of 5.8 were obtained as the wire diameter was progressively reduced by electropolishing (Fig. 6). These measurements show no definite evidence of conical texture; the total variation of α across the diameter rarely exceeds the experimental error of $\pm 1.5°$, for both the <111> and <100> components. The results reported here also show little difference in conical angle between the two texture components through the specimen. These observations contrast with the results of Schmid and Wassermann[28,29], who found that the conical angle reached a maximum of 9° for hard-drawn copper just inside the wire surface and decreased to zero at the wire centre. This inconsistency could be explained by differences in the deformation conditions used, in particular to strain-rate per pass, the design of the dies, and whether or not swaging or extrusion were used during the deformation.

Wires drawn to various strains, with 10% electropolished off their diameter, show only small conical angles between $\pm$ 1.5° i.e. mostly within experimental error (Fig. 7). Again, there is little variation between α for the <100> and <111> components. It is perhaps significant that these measurements occurred in the region just below the surface for which Schmid and Wassermann measured the largest conical angles.

The above conclusions, viz. that there is little difference in α for the <111> and <100> texture components with changing (polished) diameter or strain, is at variance with Schmid and Wassermann who reported that the <111> conical angle is much greater than the <100> conical angle[28], for copper drawn at 293K. The present results also do not support these authors' suggestion[28] that the maximum conical angle for the <111> texture corresponds to the die angle.

(ii) **Increasing Alignment of Texture**

In the X-ray diffraction patterns from drawn wires, the reflections are not as sharp as for single crystals, but are lengthened into short arcs along the diffraction circles. This lengthening occurs because the crystallites are rarely oriented exactly in the <100> or <111> directions. The spatial distributions of the crystallite directions with respect to the texture axis, as either specimen strain or thickness varied, were measured by determining the half-length ($\Delta\phi$) for the (200) reflection. Fig. 8 shows that the half-length decreases with increasing deformation, apart from a small, unexplained, local maximum near strains of 6 for both textures. Fig. 8 also shows that the crystallite alignment $\Delta\phi$ for <100> textures is always less than $\Delta\phi$ for <111> textures.

The largest changes in $\Delta\phi$ occur for the reflection corresponding to the <100> texture formed at 373K and 473K; $\Delta\phi$ decreases from 13.0° to 7.5° and 12.0° to 5.5° respectively. After deformation at 473K there was often less than 3° difference between $\Delta\phi$ for peaks corresponding to the <111> and <100> textures. However, after drawing at 293K, the difference was more in keeping with previous work[28,30], and $\Delta\phi$ for the <111> texture was approximately half that for the <100> texture.

Fig. 9 shows the variations of $\Delta\phi$, after progressive reduction of the diameter by electropolishing, for wires drawn to strains of 5.8 at 293K and 473K. There is increasing alignment of the crystallite directions to the wire axis as the centre of the wire is approached. At $d/d_0 = 0.3$ (where d_0 and d are the initial and final diameters respectively) values of $\Delta\phi$ for both texture components for the two drawing temperatures fall within 4.5° $\pm$ 0.5°. The increase in $\Delta\phi$ for $d/d_0 < 0.2$ at 473K possibly occurs because the very small amount of material examined ($d < 0.1$mm) produces a low signal-to-noise ratio at the diffractometer. Both these results show that the fibre texture in deformed copper becomes more aligned to the deformation axis with increasing deformation and closer proximity to the wire centre.

4.2 Cylindrical Texture

The (200) reflections for deformed copper include an equatorial reflection ($\phi = 0$) arising from the <100> texture, and two off-axis reflections at $\phi = \pm 35.26°$ arising from the <111> texture (Fig.3). For a true fibre texture, these latter reflections should be of equal intensity. However, most of our measurements show that the intensity of the reflection at +35.26° is approximately twice that of the other at −35.26°. If some form of cylindrical <111> texture was present, this intensity asymmetry could be attributed to differing path lengths and absorption corrections for the two reflections[31].

The above hypothesis, of a part cylindrical texture in the drawn copper, was confirmed using pole diagrams for two different specimens prepared from wires. The first specimen consisted of a

raft of parallel copper wires (strain 6.2, wire diameter 0.25mm) cast into epoxy resin, with the leading end of each wire during drawing kept to the same end of the raft (Fig. 10(a)). The raft was then mechanically and chemically polished down to a diametral section. Both the (111) and (200) pole figures showed the presence of <111> and <100> textures (Fig. 11) not, however, as "pure" fibre textures but with some component of a cylindrical or single-crystal type texture.

The above result does not enable a distinction to be made between the two latter possibilities of texture type A second specimen was prepared in which a central cylinder was spark-machined from a 2.67mm diameter wire (strain 1.55), leaving a 0.25mm thick tube which could be slit and opened flat (Fig. 10(b)). For cylindrical texture, this flattened annulus should possess an overall crystallographic orientation similar to that of a rolled sheet, viz. <111> and <100> directions in the direction of drawing, and another preferred direction perpendicular to the sheet (i.e. in the original radial direction). For single-crystal type texture, the sheet specimen should not possess any major orientation perpendicular to the sheet, but only the predetermined wire axis orientations. Fig. 12 shows the pole diagram from a sheet specimen, which is consistent with an intermediate cylindrical texture. Figs. 11 and 12 both show slightly more cylindrical texture for <111> than for <100> texture.

4.3 Texture Formation

During deformation, the initial random texture (strain = 0) transforms into definite texture components at strains of 1.6, as shown by the absence of diffracted X-rays in regions between maxima on the Debye-Scherrer rings. For strains of 1.6 and greater, the diffraction peaks corresponding to the <111> and <100> textures are so strong that the entire wire can be considered to be in these textures only.

The predominance of the texture components is described by

$$\%\,\langle 100\rangle = \frac{[\text{corrected intensity due to } \langle 100\rangle]}{[\text{corrected intensity due to } \langle 100\rangle + \text{corrected intensity due to } \langle 111\rangle]} \times \frac{100}{1}$$

The observed percentages, for drawing temperatures in the range 293K to 473K, plotted against drawing strain approximately follow straight lines which intercept the ordinate between 46% and 49% (Fig. 13); the lines shown are for "least-squares" fits. This intercept is consistent with an initial random orientation for which the intercept is 50% by definition. The unstrained, 5.7mm diameter, polycrystalline rod was shown to be randomly oriented by a uniform diffraction line intensity.

The %<100> in drawn copper varies from 46% to 54% at a strain of 1.6 to between 35% and 88% at a strain of 7, depending on deformation temperature (Fig. 13). There is no evidence to suggest that the %<100> has reached its limiting values at a strain of 7; this result is similar to that for drawn gold[5], but is unlike the results obtained at similar temperatures for silver[6], which show 100% <100> for strains above 5.

4.4 Effect of Thermal Annealing

Evidence that the cube orientation forms preferentially in copper during recrystallisation was obtained by determining the textures present after annealing several deformed wires at 573K for two hours. The annealing resulted in wires with more than 90% by volume in <100> orientation along the wire axis, for wires deformed at 293K and 473K. This result is consistent with the H.V.E.M. work of Hutchinson[20] who found that copper foils recrystallised with a preferred cube orientation.

5. DISCUSSION

5.1 Mechanisms of Texture Formation in Copper

It is difficult to compare the present texture results strictly with those available in the literature. However, our results show a higher percentage of <100> texture than previously reported for copper[8]. Results surveyed by Ahlborn and Wassermann[8] indicated between 10% and 40% <100> in copper deformed at room temperature, except for the results of Freda and Cullity[1], whose initial material contained nearly 100% <100>.

In the present work, the percentage of <100> texture at a given strain increases with temperature (Fig. 13). Any mechanism of <100> texture formation based on twinning is thus

inappropriate here, since the incidence of twinning in f.c.c. metals decreases as temperature increases. This conclusion is consistent with the absence of X-ray reflections from twins (i.e. <115>) in specimens deformed between 293K and 473K. Any mechanism postulated must encourage an increase in <100> texture with increasing temperature of deformation. The temperature dependence of cross-slip is not consistent with the increasing percentage of <100> texture with increasing deformation temperatures beyond room temperature. It is concluded therefore that cross-slip is not the dominant mechanism responsible for the texture formed at 373K and 473K in the copper wires examined.

At 293K the percentage of <100> texture decreases linearly with increasing strain, while the percentage <111> increases correspondingly.

These variations are consistent with a recovery mechanism in which dislocation cross-slip operates at temperatures around 293K to maintain a predominantly <111> deformation texture. Evidence that a recovery mechanism operates in copper at 293K is found in the decreasing rate of work hardening shown by copper deformed at 293K[32]. It follows that, if this interpretation is correct, another mechanism must become predominant above 293K, leading to an increasingly <100> texture with increasing temperature. Since this other mechanism supplants a recovery process and is apparently not based on slip or twinning, we next consider the effects of recrystallisation on texturing in copper.

5.2 Cube Texture in Copper

In our tests, annealing deformed wires at 573K resulted in recrystallisation which left more than 90% by volume of the wires in <100> texture. Thus, it seems reasonable to assume that partial recrystallisation during large deformation by drawing at temperatures above 293K would produce <100> textured crystallites. Since the amount of recrystallised material would increase with both temperature and drawing strain, this explanation is qualitatively consistent with the observed relationships (Fig. 13).

Evidence from transmission electron-microscopy of the wires[33] supports the proposed interpretation. Here, many small, dislocation-free nuclei are observed, together with dislocation cells forming within subgrains; these subgrains are, in turn, surrounded by boundaries similar to those formed during recrystallisation[33] (Fig. 14). These small subgrains are found in wires drawn at temperatures from 196K to 473K, but the probability of their occurrence increases with temperature of deformation[33]. There is thus evidence of increasing recrystallisation in the deformed structure as the temperature is increased.

It should be noted that the <100> orientation in copper has been found to be stable in that no new textures form after extensive further deformation [10]. This stability was shown by swaging single crystals of copper to 96% reduction in area and measuring the texture components; crystals oriented near <100> tended to move closer to <100> and those near <111> moved closer to <111>. Results of a less comprehensive study for aluminium were consistent with those for copper.

The explanation that the <100> texture results during a recrystallisation process is also consistent with the results shown in Fig. 8, viz. the crystallite alignment for <111> texture is always greater than for <100> texture. This result probably arises because the <111> texture consists of recovered crystallites which have been subjected to more deformation, thereby promoting closer alignment of texture with the wire axis, than the <100> texture which arises by recrystallisation. An alternative explanation for decreased <100> crystallite alignment is that the newly nucleated <100> crystallites show a relatively large spread of orientation, but this deviation from true <100> texture direction does not prevent recrystallisation of that crystallite proceeding.

5.3 Aluminium Texture

The case of texturing in aluminium is relevant to our proposed explanation for drawn copper since softening processes operate in aluminium at temperatures close to room temperature, (homologous temperature $T/T_m \sim 0.31$, c.f. copper at 473K with $T/T_m \sim 0.35$). Thus it might be expected that the <100> texture should predominate in deformed aluminium. In fact, the reverse is the case; 92% <111> and 8% <100> texture is found in aluminium wires drawn to a strain of 4.2 at room temperature[8].

Thermal annealing of cold-rolled aluminium sheet shows two main textures[34,35,36] viz. an "R" texture[37] associated with (123) $[1\bar{2}1]$ or (112) $[11\bar{1}]$ and a cube texture. The extent to which the cube texture predominates is very much a function of the temperature of annealing and

the time taken to reach that temperature; more rolling texture is retained when aluminium is heated more slowly to the annealing temperature[35]. This result is consistent with the earlier discussion (sect. 2.4) in which evidence was presented showing that recovery processes in aluminium occur to quite high homologous temperatures. Hence, it appears that the <100> texture does not predominate in drawn aluminium because at most temperatures of deformation recovery maintains the <111> deformation texture. Simultaneously, the internal energy is reduced to such an extent that recrystallisation, which gives rise to increasing cube orientation, is unlikely to predominate until temperatures much higher than room temperature.

5.4 Cube Texture Formation

As yet there is no definitive explanation for the preference of f.c.c. metals to recrystallise into a cube orientation, although the mechanism by which recrystallisation occurs has as its basis either an oriented nucleation or oriented growth mechanism[20,38]. The precise mechanism is still subject to uncertainty. Nevertheless such mechanisms will operate so as to lower the internal energy of the deformed material.

It is relevant here that in wires heavily drawn at homologous temperatures below 0.5, the cells and grains usually become highly elongated in shape. Hence, during further deformation each cell undergoes a similar strain to each of its neighbours. In this case cells with the highest elastic modulus will have the highest elastic internal energy; the elastic energy is proportional to modulus under conditions of equal strain.

The elastic moduli of f.c.c. metals are quite anisotropic. For copper the maximum-to-minimum ratio is 2.83 and rather less for aluminium (1.2); values for other metals are tabulated in Table 1. The maximum and minimum values for Young's modulus occur in the <111> and <100> directions respectively. Stuwe[18] maintains that, in f.c.c. metals, the free energy is a maximum when the cube axes coincide with the directions of principal stress. Although this concept is valid for "constant-stress" conditions on each cell, "constant-strain" on each cell is applicable for highly elongated cells, as in the present wires[33]; here the elastic energy is a minimum for cube oriented cells.

If minimising the elastic energy of each cell was the factor determining which grain or cell should nucleate or grow, the <100> texture would predominate. However the temperature at which predominance occurs varies from metal to metal, depending on the temperature-dependence of their softening processes. It is only when recrystallisation is favoured over recovery that reorientation into <100> becomes common. This conclusion is consistent with the cube orientation in aluminium not becoming predominant until higher temperatures[24]. It should be noted that both recovery and recrystallisation are thermally activated processes; thus the driving force for recrystallisation is aided by a large elastic anisotropy. This large anisotropy will therefore assist in the lower temperature recrystallisation of copper compared with aluminium.

Table 1.

Metal	Young's Modulus Maximum-to-minimum ratio	Reference
Al	1.2	39
Ni	1.32	40
Ag	2.68	39
Au	2.7	39
Cu	2.83	39

6. CONCLUSIONS

The deformation textures of drawn copper wires have been investigated using X-ray diffraction techniques. It is concluded that:

(a) Two textures, viz. <100> and <111> occur. The relative amounts of material in the textures depend on the temperature of deformation and strain.

(b) The predominance of the <111> texture at 293K is ascribed to the dominance of recovery, over recrystallisation, thereby maintaining the texture formed by deformation.
(c) As deformation temperatures increase between 293K and 473K, the increasing predominance of the <100> texture is explained by the growing preference for recrystallisation to occur rather than recovery.
(d) The preference for recrystallisation to result in the formation of a cube texture is attributed to the anisotropic elastic internal energies being minimised by recrystallisation into the cube orientation.
(e) There was no evidence that twinning plays any role in texture formation in copper deformed at temperatures between 293K and 473K.
(f) Reports[28,29] that a texture with conical angles of up to 9° is formed during wire drawing were not confirmed. Conical angles were usually less than ± 1.5° and within experimental error.
(g) The larger spatial distribution of crystallite orientations for <100> textures compared with <111> textures could be caused by the smaller deformation undergone by the recrystallised <100> areas compared with the recovered <111> areas, or by the spread of orientations of the nuclei recrystallising near to the cube orientation.
(h) The textures formed are neither ideally "fibre" nor completely "cylindrical" in type, but appear to be intermediate in character between these types.

ACKNOWLEDGEMENTS

The authors wish to thank Mr. M.D. Engellenner for his assistance in obtaining the diffraction results and in preparing diagrams. We also wish to acknowledge the helpful translations from German provided by Dr. M.E. Packer.

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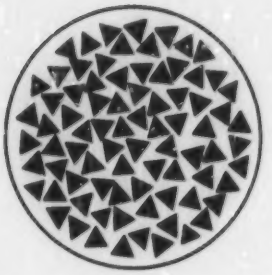
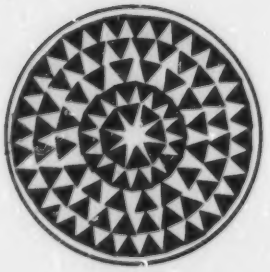
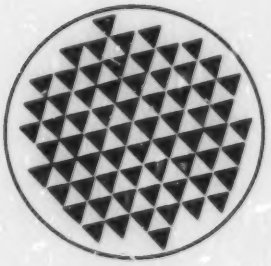


(a)



(b)



(c)

Fig. 1 Schematic representation of crystallite orientation in wire cross-section for the three kinds of texture in f.c.c. wires.

(a) fibre
(b) cylindrical
(c) single crystal.

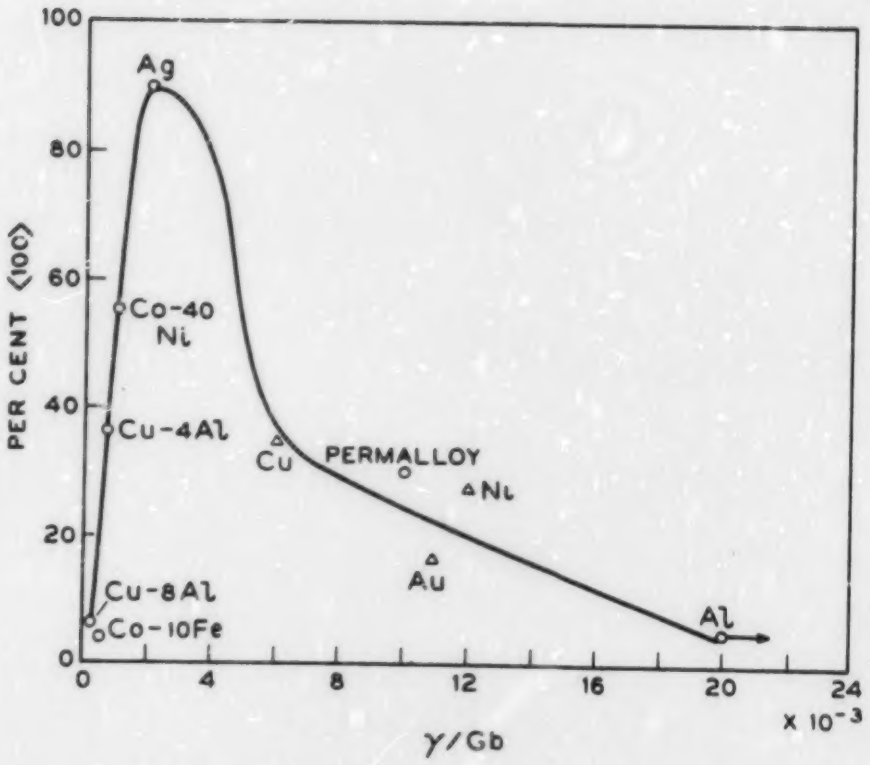


Fig. 2 Correlation between the percentage of (100) texture and the stacking-fault-energy parameter (γ/Gb) [7], in heavily drawn wires of f.c.c. metals, after English and Chin[7].

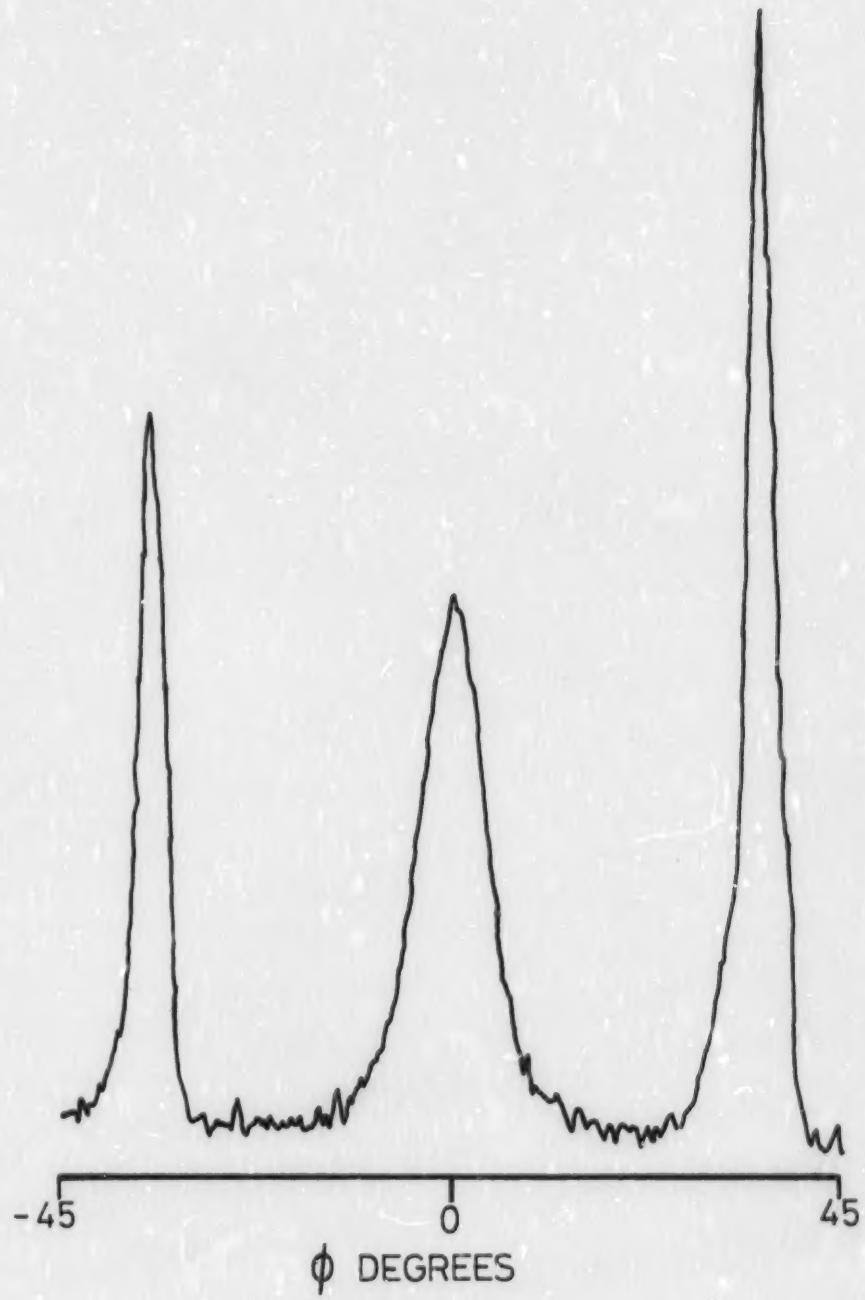


Fig. 3a Typical computer-corrected intensity distribution along the Debye-Scherrer arcs for drawn copper wire. Temperature of deformation 293K, strain = 3,6;(200 reflection).

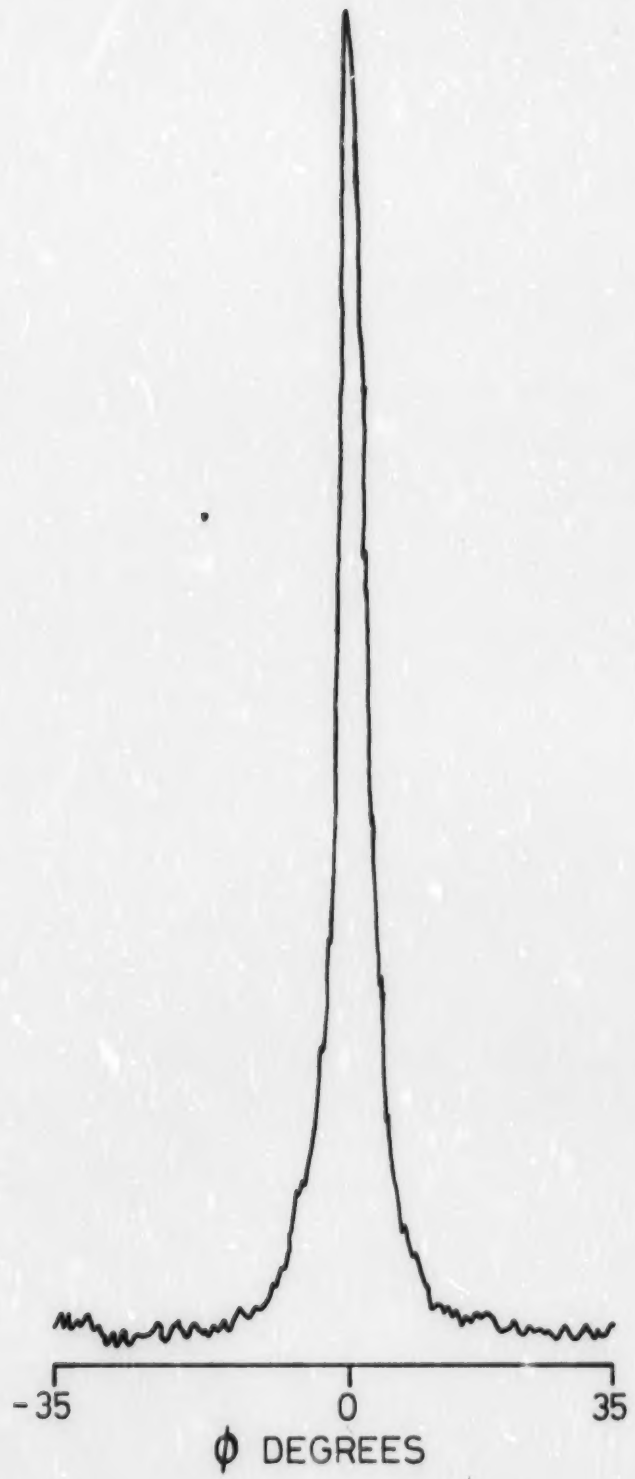


Fig. 3b Typical computer-corrected intensity distribution along the Debye-Scherrer arcs for drawn copper wire. Temperature of deformation 293K, strain = 3,6;(220 reflection).

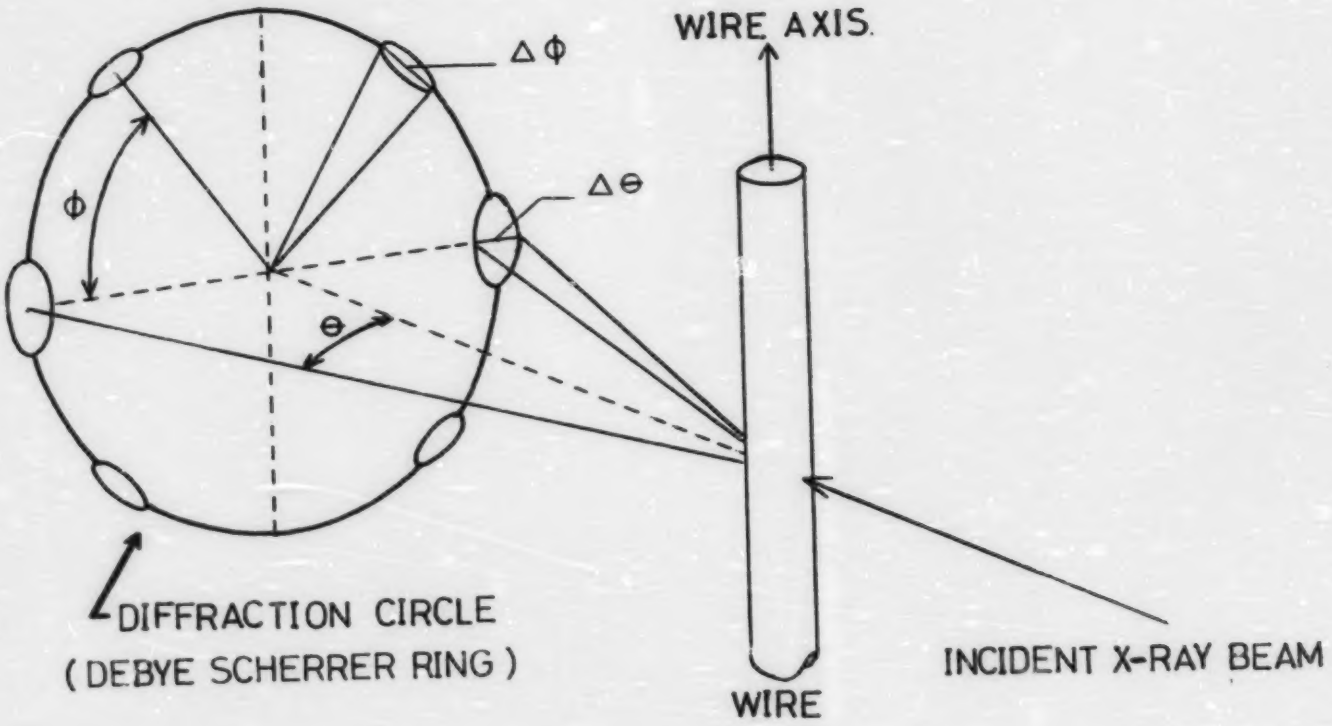


Fig. 4 Schematic arrangement during X-ray diffraction measurements. The circumferential arc length ($\Delta\phi$) is caused by randomness of crystallite orientation about the main texture direction. θ is the Bragg angle.

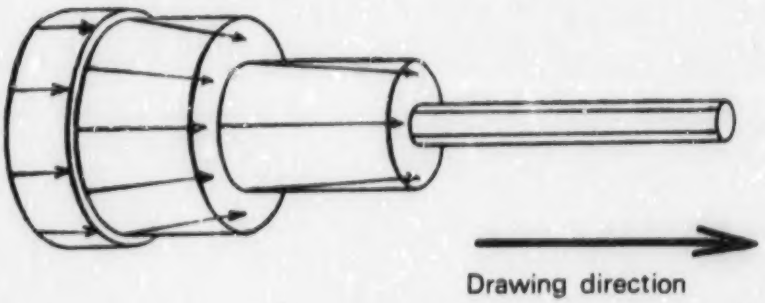


Fig. 5 Schematic representation of the conical texture of hard drawn copper wire as proposed by Schmid and Wassermann[28]. The directions of the fibre axes and degree of preferred orientation are represented by the directions and lengths of the arrows.

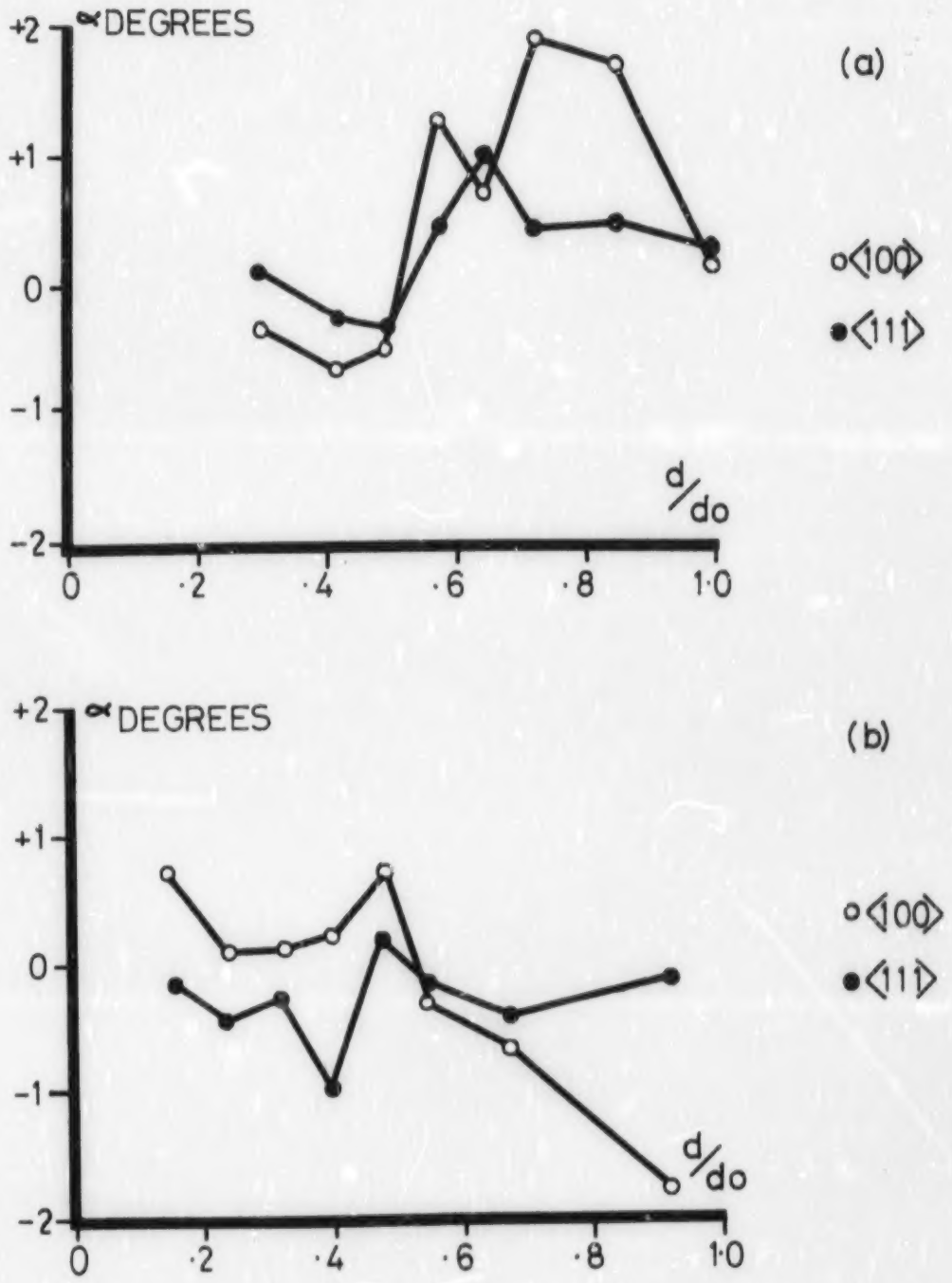


Fig. 6 Variation of the conical angle (α) for copper (strain = 5,8) drawn at (a) 293K and (b) 473K, as the wire diameter (d) is progressively reduced by electropolishing. d_o is the original wire diameter.

The angles plotted for the ⟨111⟩ texture were obtained by measuring the displacement of the peak near 35,26° from its equilibrium position.

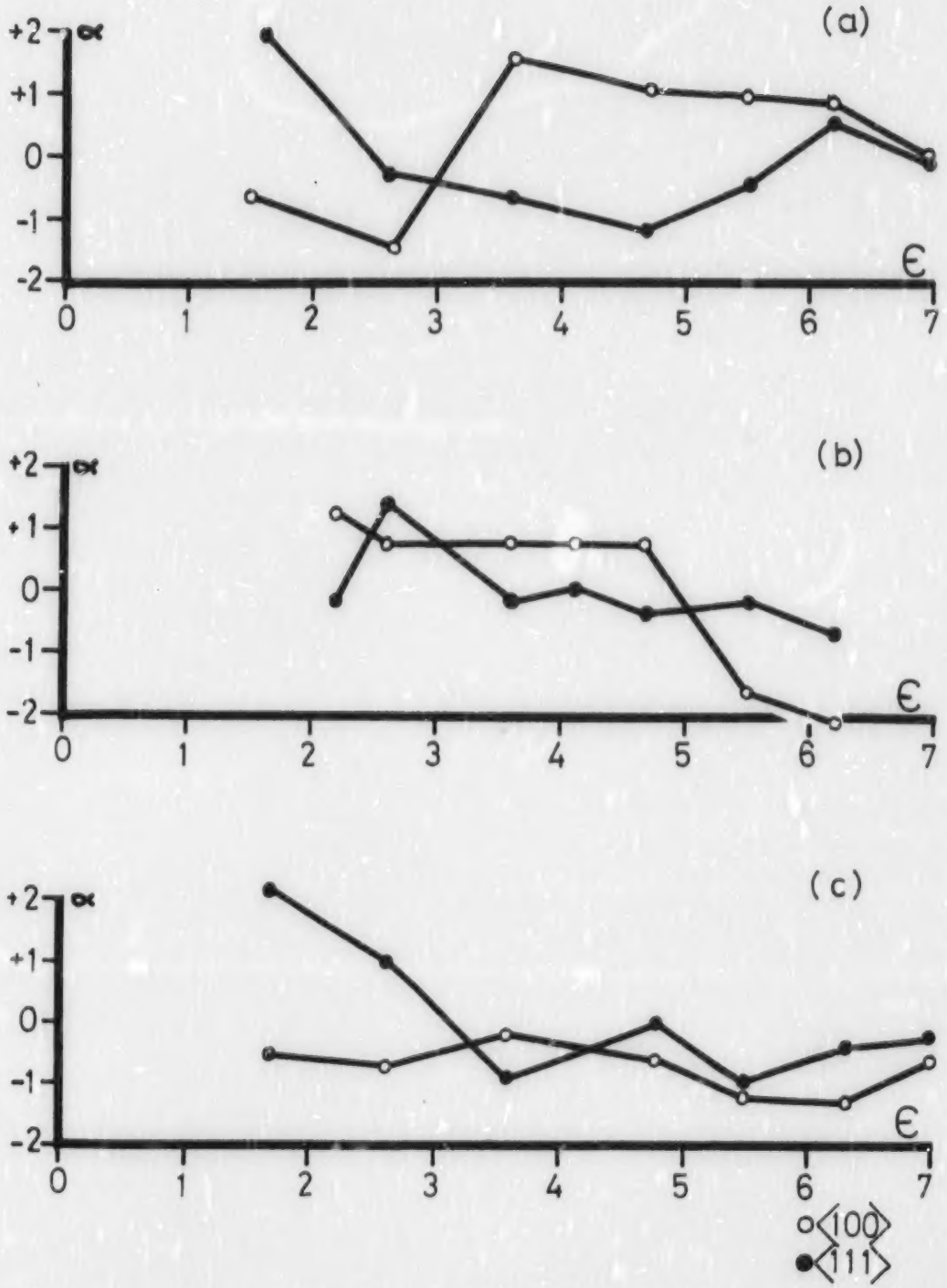


Fig. 7 Variation of the conical angle (α) for copper drawn at (a) 293K, (b) 373K and (c) 473K, as a function of strain (ϵ). The angles plotted for the ⟨111⟩ texture were obtained by measuring the displacement of the peak near 35,26° from its usual position.

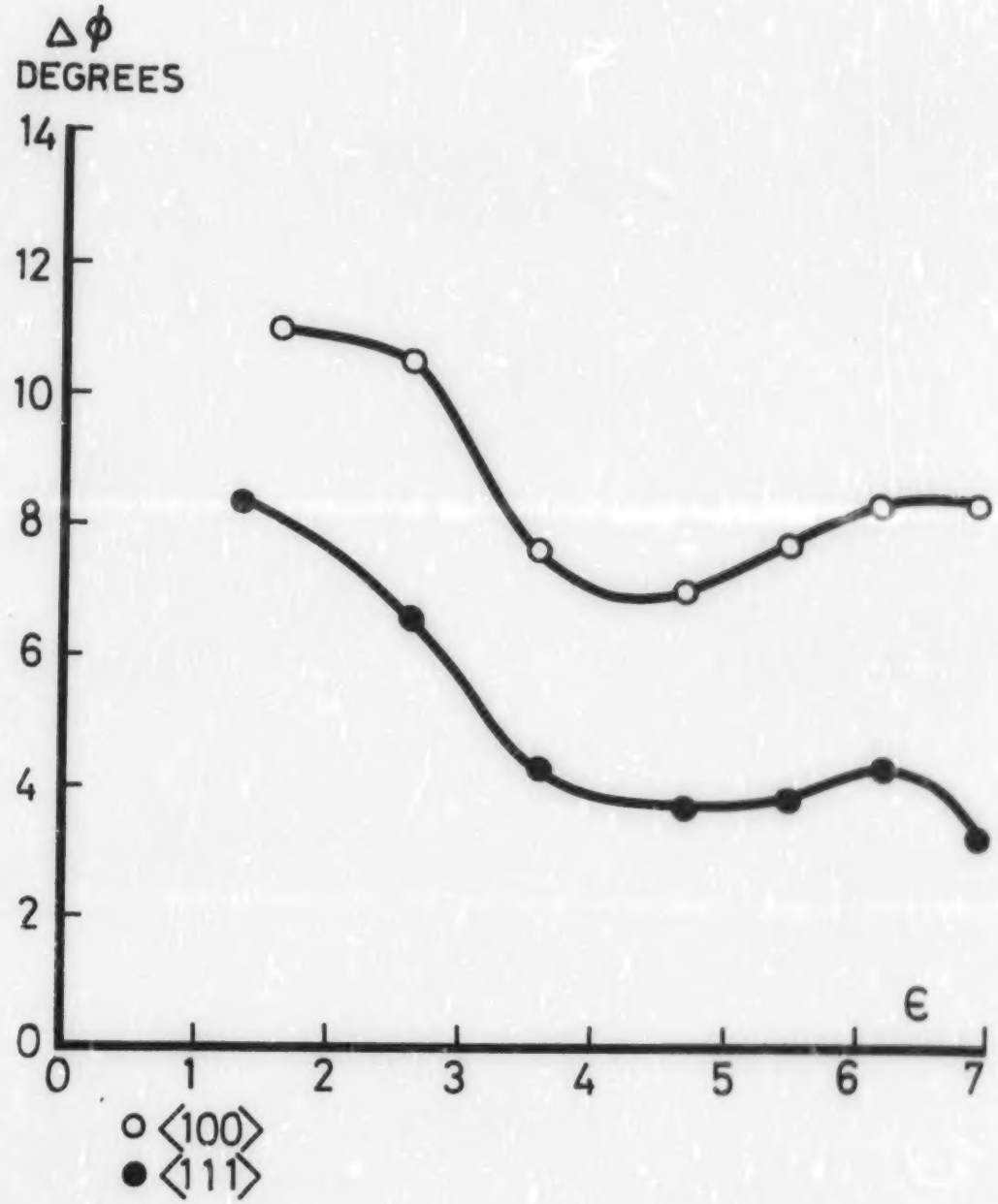


Fig. 8a Variation in crystallite alignment ($\Delta\phi$) with strain (ε) for copper deformed at 293K. $\Delta\phi$ values were taken from the peaks of the (200) reflection.

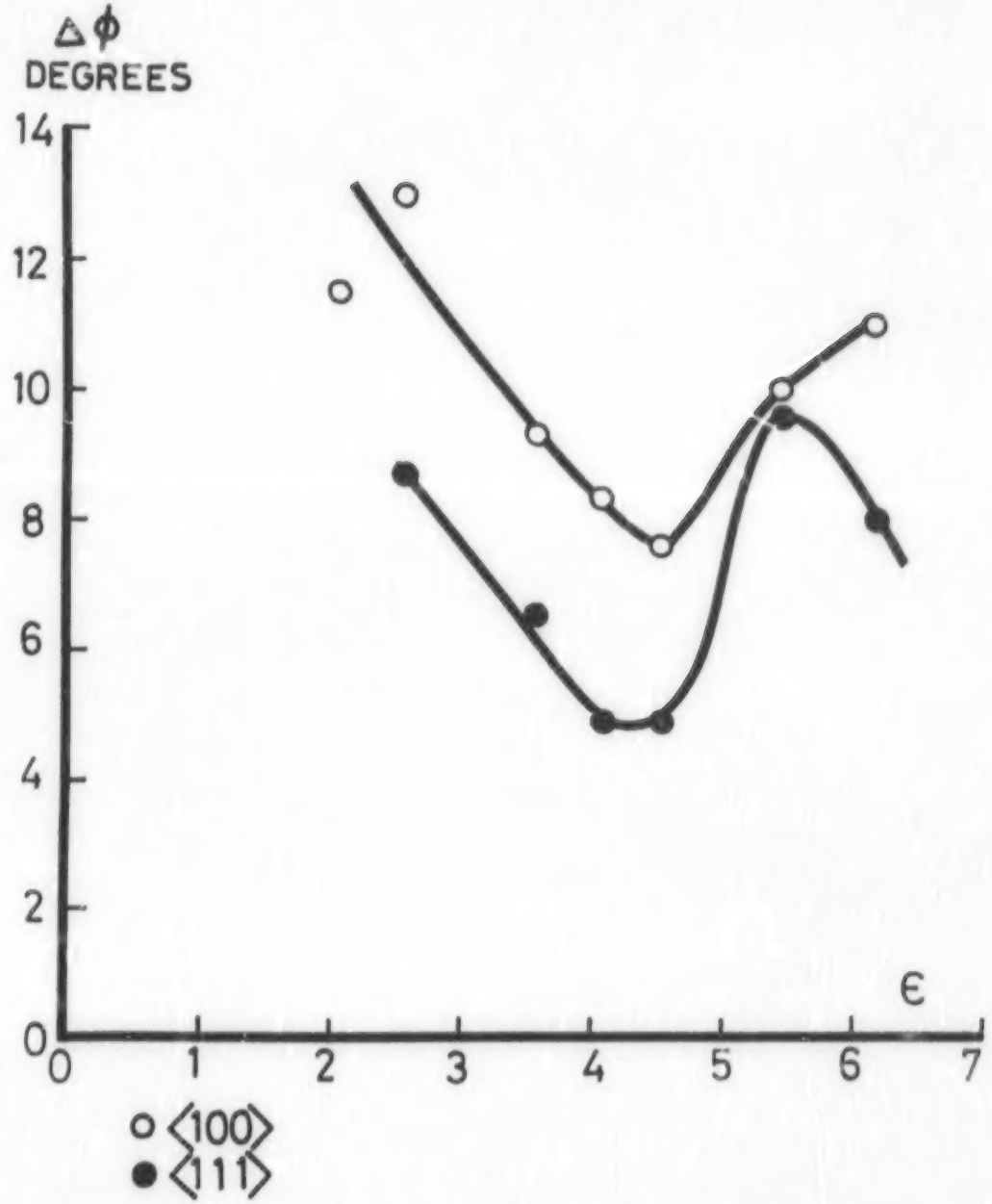


Fig. 8b Variation in crystallite alignment ($\Delta\phi$) with strain (ϵ) for copper deformed at 373K. $\Delta\phi$ values were taken from the peaks of the (200) reflection.

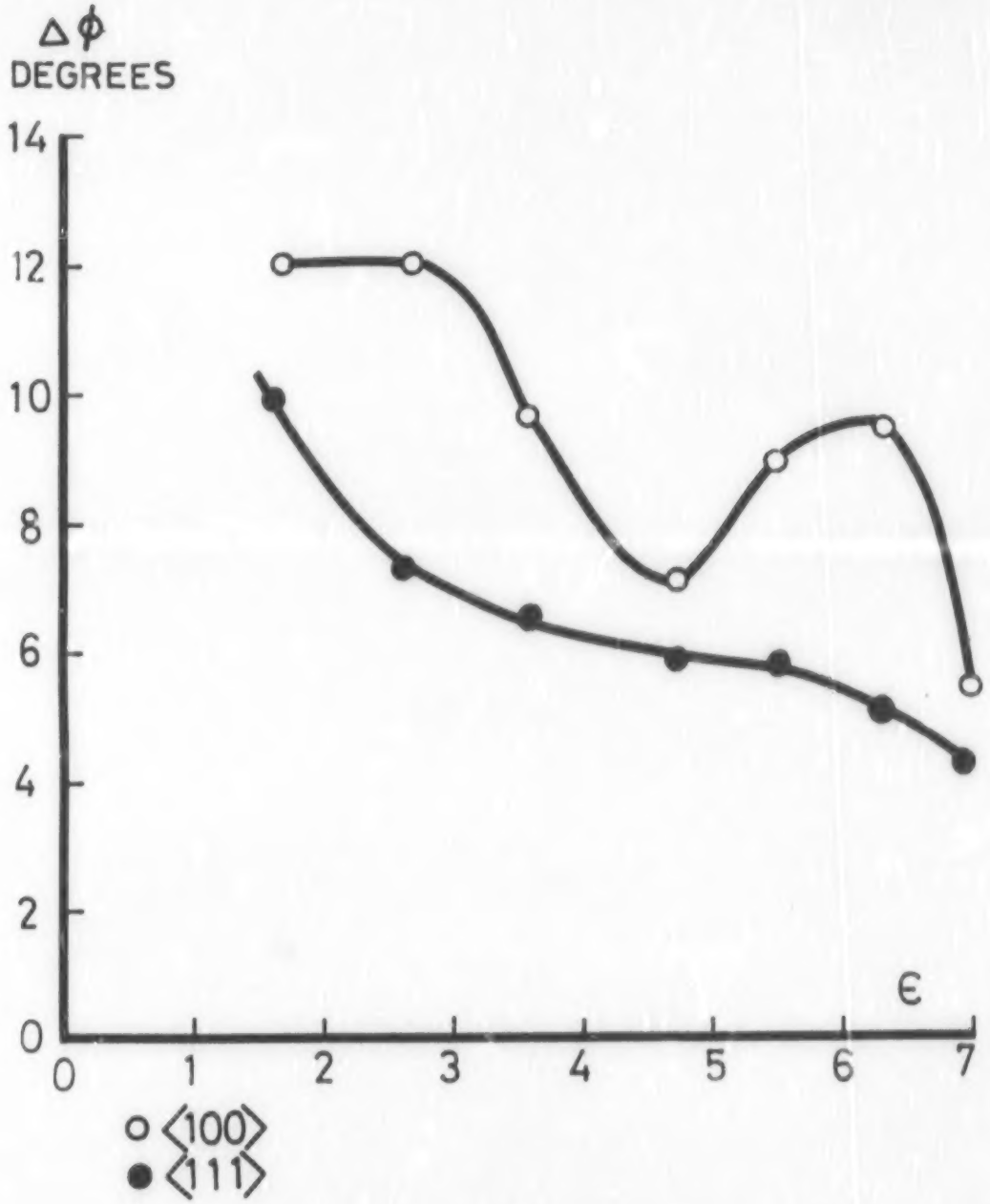


Fig. 8c Variation in crystallite alignment ($\triangle\phi$) with strain (ϵ) for copper deformed at 473K. $\triangle\phi$ values were taken from the peaks of the (200) reflection.

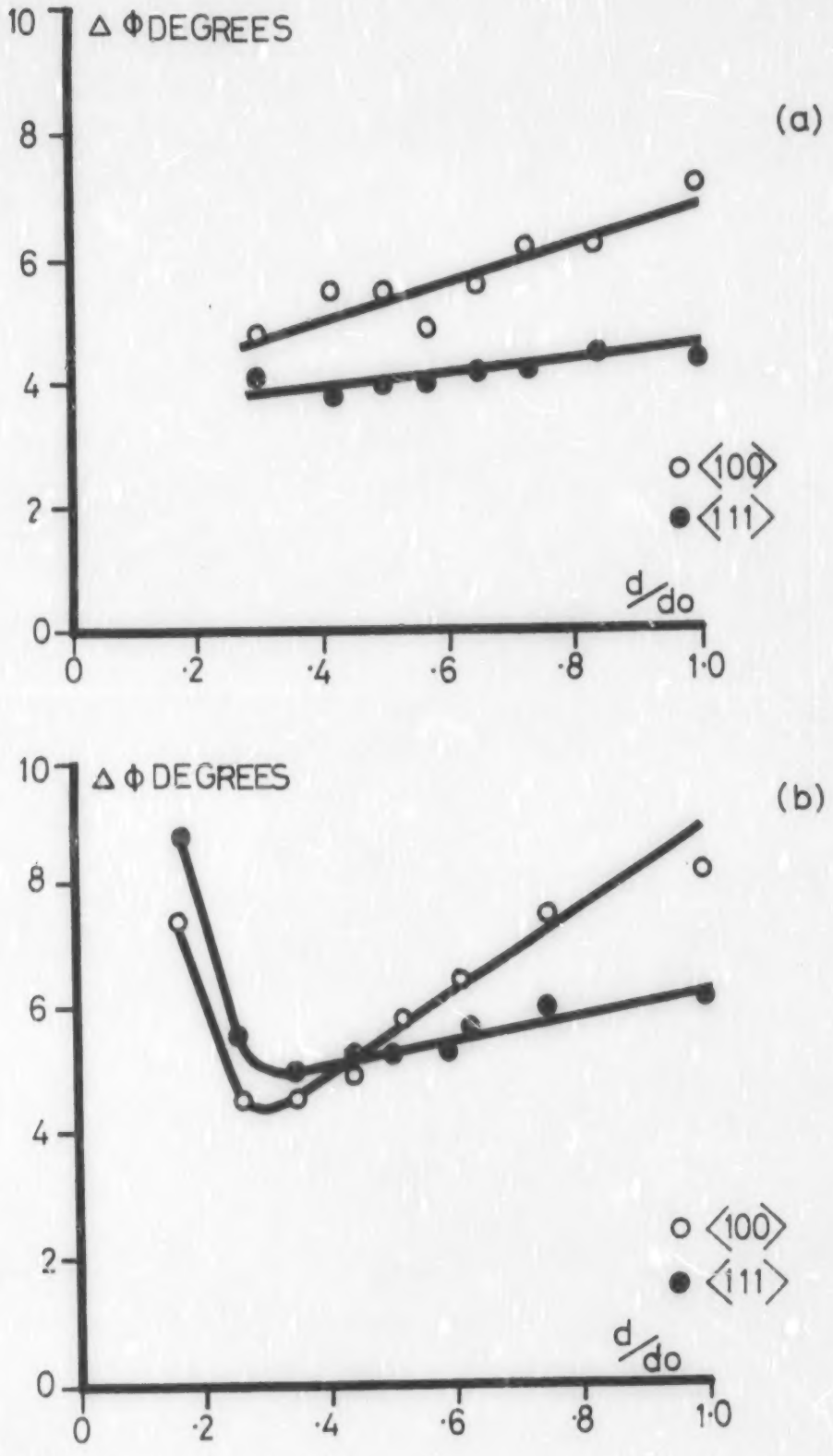
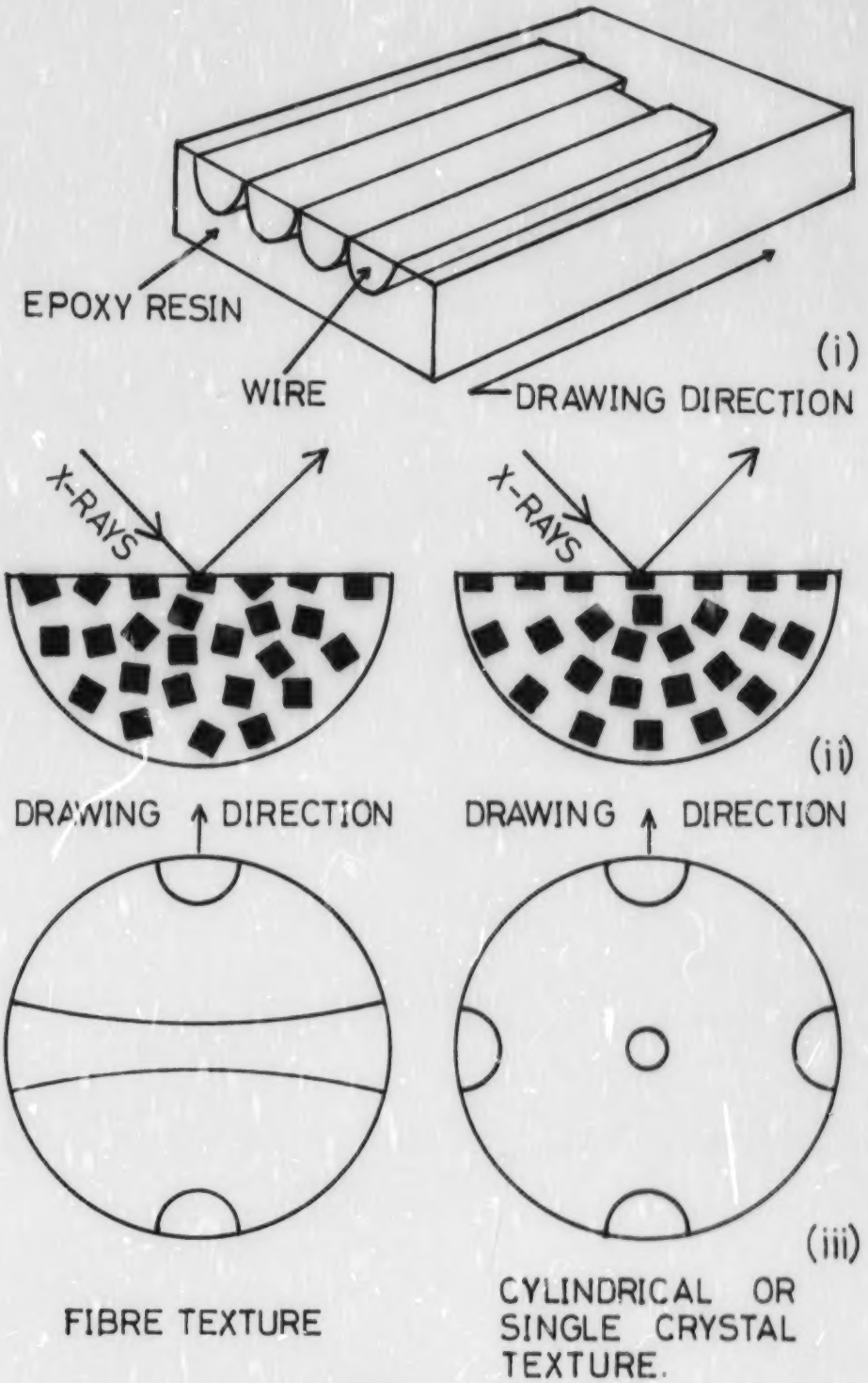
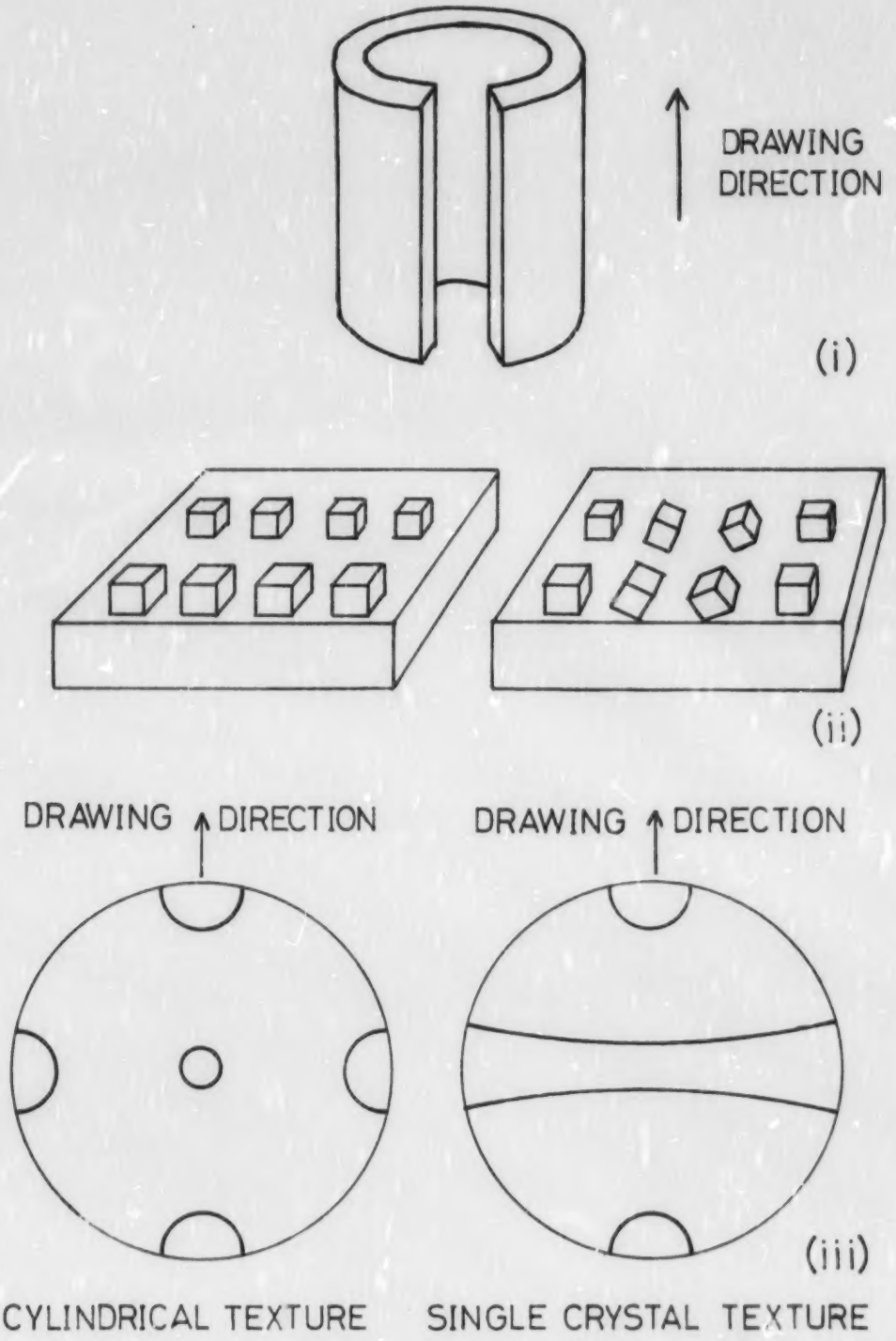


Fig. 9 Variation in crystallite alignment ($\triangle\phi$) as the wire diameter (d) is progressively reduced by electropolishing. Copper deformed at (a) 293K and (b) 473K to a strain of 5,8.d_o is the original wire diameter.



(a) "RAFT" TYPE SPECIMEN

Fig. 10a Schematic arrangement of the specimen used in distinguishing between fibre texture, cylindrical texture and single-crystal texture, and the form of ideal pole diagrams for each texture type. "Raft" specimen.
(i) Cast wires (ii) Schematic cross section (iii) Typical (200) pole figure



(b) "SHEET" TYPE SPECIMEN

Fig. 10b Schematic arrangement of the specimen used in distinguishing between fibre texture, cylindrical texture and single-crystal texture, and the form of ideal pole diagrams for each texture type. "Sheet" specimen.
(i) Original split annulus (ii) Flattened annulus (iii) Typical (200) pole figure

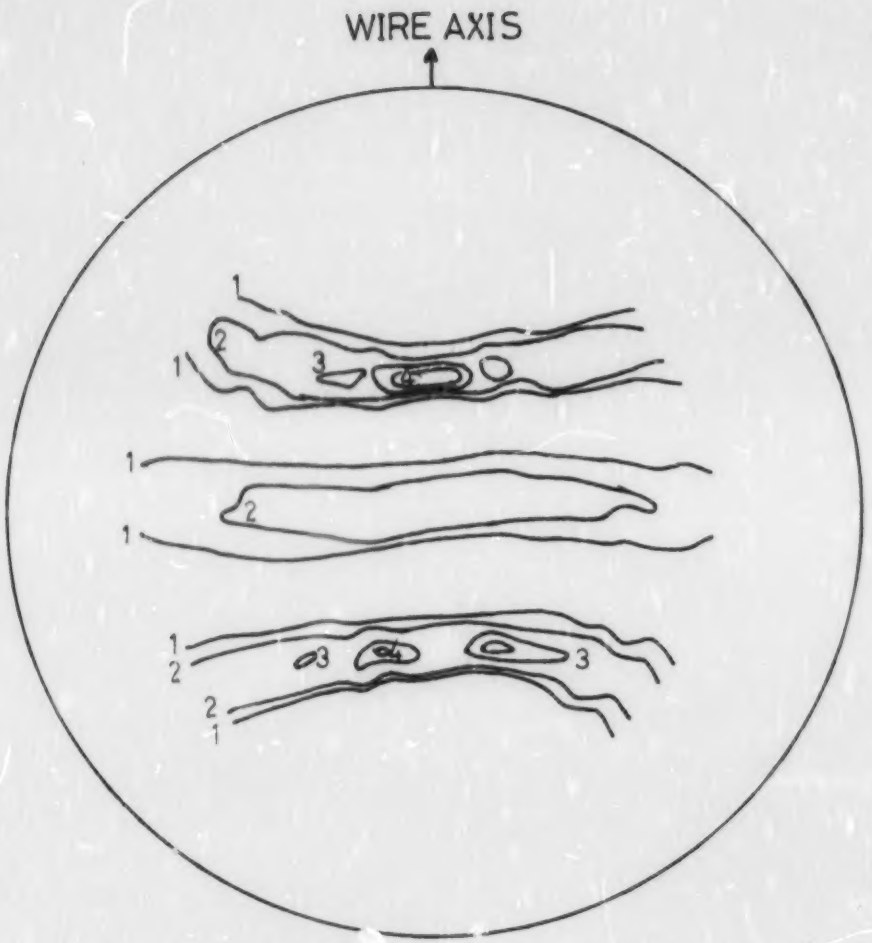


Fig. 11a Pole figure obtained using "raft" specimen. Copper deformed to a strain of 6,2 at 293K;(200) pole figure.

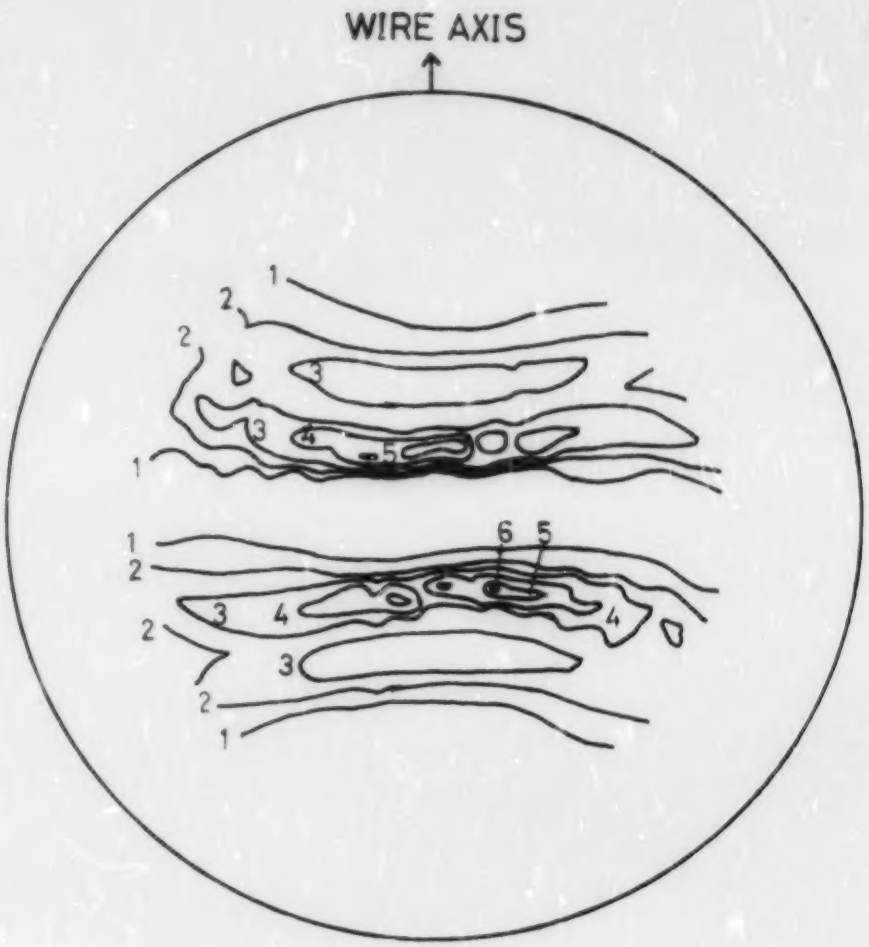


Fig. 11b Pole figure obtained using "raft" specimen. Copper deformed to a strain of 6,2 at 293K;(111) pole figure.

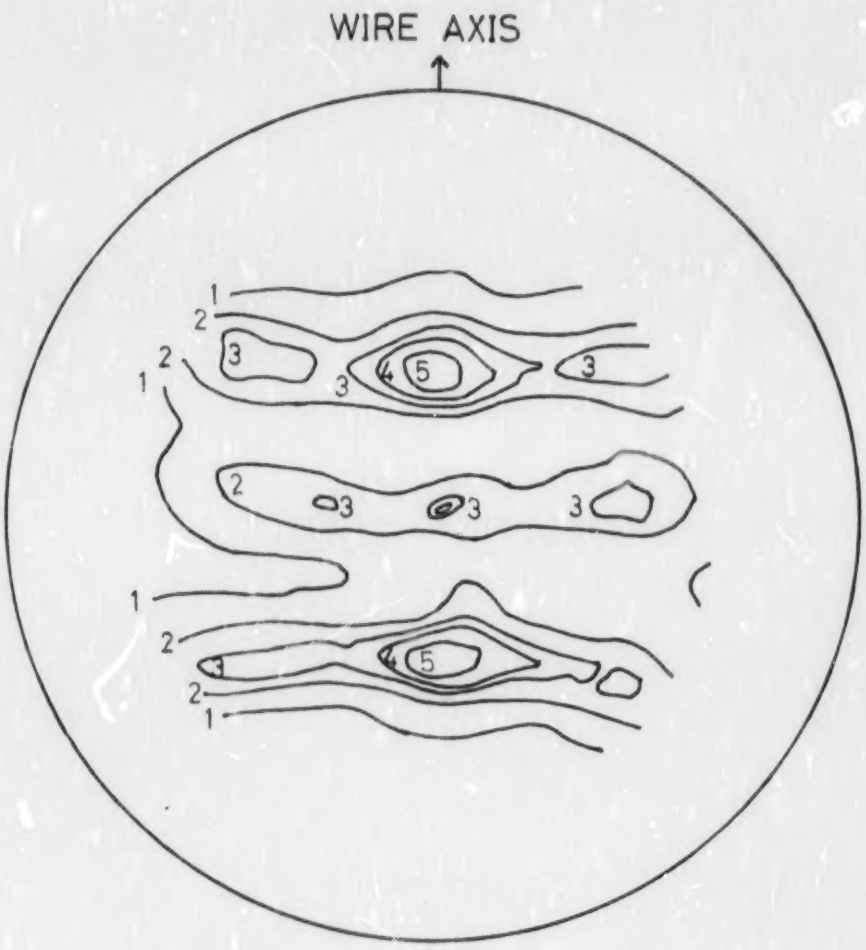


Fig. 12a Pole figure obtained using "sheet" specimen. Copper deformed to a strain of 1,6 at 293K;(200) pole figure.

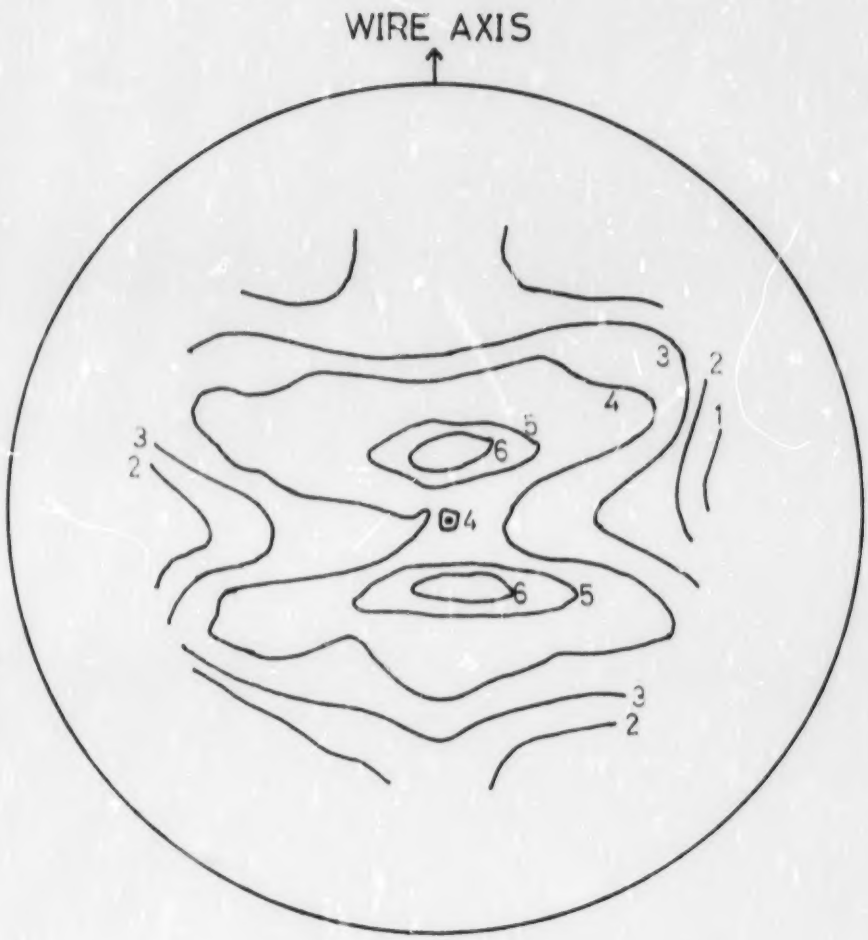


Fig. 12b Pole figure obtained using "sheet" specimen. Copper deformed to a strain of 1,6 at 293K;(111) pole figure.

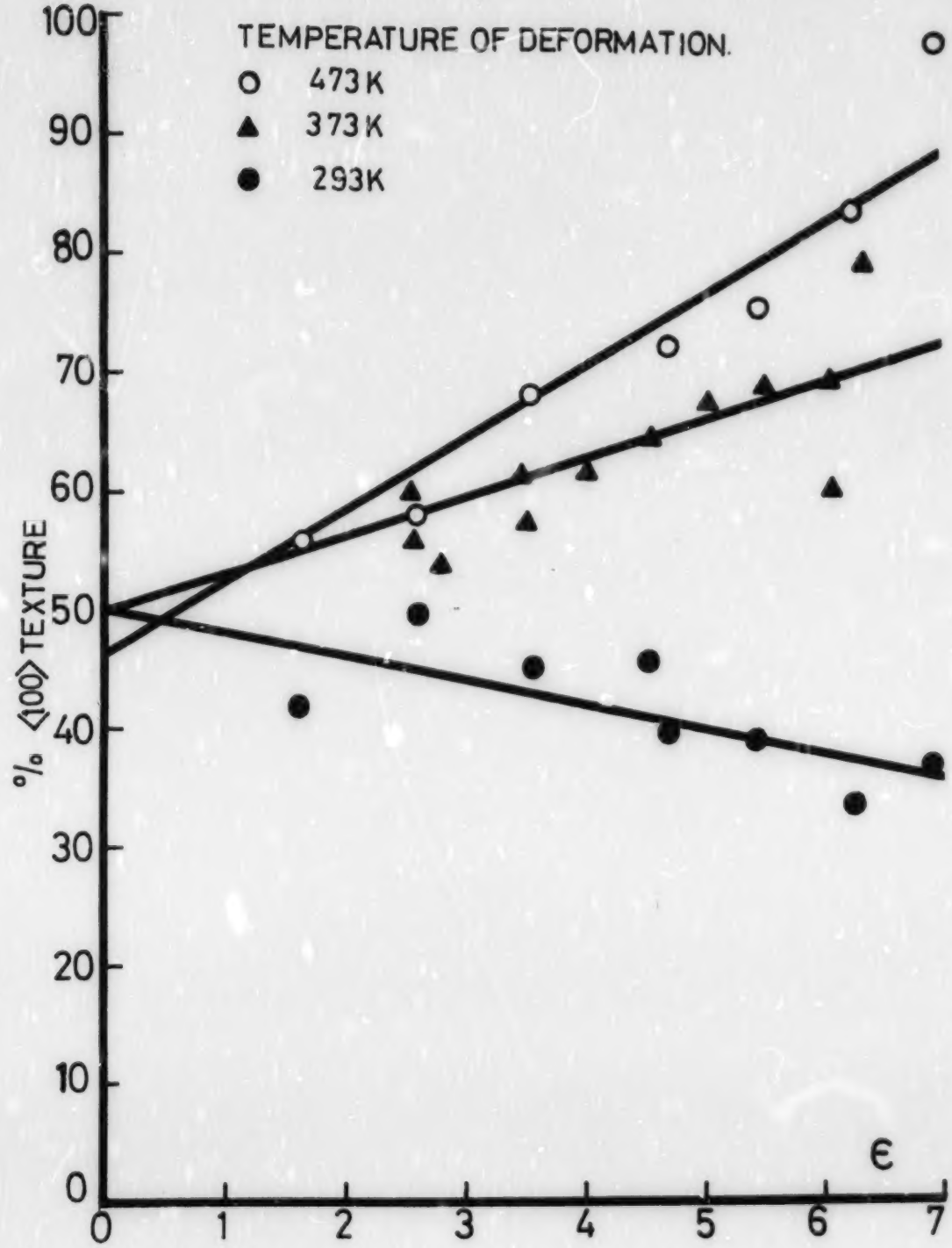


Fig. 13 Variation of the percentage ⟨100⟩ texture as a function of strain (ϵ) for drawn copper wire, at three temperatures of deformation.

Fig. 14 Typical dislocation microstructure for drawn copper (strain = 3,6), temperature of drawing 473K, section transverse to the direction of drawing[33]. Micrograph shows dislocation sub-boundaries, contained within more perfect grain boundaries.

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5. Personal Authors R.B. Nethercott and R.A. Coyle 6. Corporate Author(s)	
7. Abstract (See Reverse)	
8. Computer Program(s) — Titles and Language Not Applicable	
9. Descriptors Textures Copper Deformation X-Ray Diffraction	11. Cosati Classifications 1407, 0702, 2006 12. Task Reference (RD/P) RD 73 Task 50/7
10. Library Distribution (Defence Group) A.R.L. Library Central Library S.T.I.B. Library W.R.E. M.R.L. (Vic. N.S.W. S.A.) Central Studies Est. The Bridges Library, Duntroon	13. Sponsoring Agency Reference 14. Cost Code 7312
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ABSTRACT

An X-ray study of the texture produced by wire drawing of pure copper indicates that a double <100> + <111> texture is always found. At a deformation temperature of 293K, the predominance of the <111> texture is ascribed to the dominance of recovery over recrystallisation processes, thereby maintaining the <111> deformation texture. At deformation temperatures of 373K and 473K, the <100> texture steadily increases and the <111> texture progressively decreases. This change is ascribed to the growing dominance of recrystallisation into the cube orientation over recovery. Anisotropy of elastic moduli is used to explain the preference for the cube orientation to form during recrystallisation.

The perfection of formation of the two texture components was investigated as a function of both deformation and depth below the wire surface. The <111> component is always more closely aligned with the wire axis than the <100> component. There was no evidence that a conical texture is formed during drawing.

The textures formed are neither ideally fibre nor completely cylindrical in type, but appear to be intermediate in character between these types.